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PREDICTING RELAXATION IN STRAINED EPITAXIAL LAYERS

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Abstract

Strained epitaxial semiconductor layers, much thicker than the critical thickness, have been used as "strain-relief" buffer layers for many years. The most successful structure developed so far dates back to the 1960's, and consists of a very thick ($\sim 30\ \mu m$) layer in which the misfit is gradually and continuously increased. These structures relax completely and have a sufficiently low threading dislocation density to allow a device structure to be grown on top. This process requires a very high growth rate to produce the buffer layer in a reasonable time, which is only provided by hydride vapour-phase epitaxy. Recently, there has been interest in developing thinner structures using both graded and constant composition buffer layers, which, if successful, would resolve this problem. Here, we consider the mechanisms of strain relaxation, paying special attention to the changes in threading dislocation density and surface roughness that occur during misfit relief. An extensive series of experiments shows that the relaxation of constant composition layers, although not following current theoretical models, does appear to follow a simple empirical law. This result suggests an approach which can be used to predict the state of strain in any epitaxial structure, allowing more efficient strain-relief buffer layers to be designed.

Key Words: Heteroepitaxy, strained layers, misfit dislocations, strain relief, dislocation multiplication, dislocation blocking, $In_xGa_{1-x}As/GaAs$.

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Introduction

Strain-relief buffer layers, which bridge the lattice parameter gap between a device and a substrate, have been in use for many years in light-emitting diode (LED) structures. The conventional design, invented in the 1960's, uses a very thick buffer layer ($\sim 30\ \mu m$) with a lattice parameter that changes linearly between substrate and device (e.g., Abrahams *et al.*, 1969, 1975). These structures relax completely (within the limits of detection), allowing a strain-free device to be grown on top. Because of the size of the buffer layer, this technique has, not surprisingly, only been used in hydride vapour-phase epitaxy systems, which can grow the structure in an hour or so. The more modern deposition techniques, such as molecular beam epitaxy (MBE) and metal-organic vapour-phase epitaxy (MOVPE) have much lower growth rates, and would typically require up to a day to produce the same structure. If this strain relief buffer technique is to be applied to more recent device designs, such as the semiconductor laser, it would be advantageous to grow both buffer layer and device in the same growth run. Until recently, the majority of theory and experiment has been concerned with predicting and measuring the "critical" epitaxial layer thickness for the onset of dislocation generation in strained layers. Here, we are primarily concerned with understanding the strain relief processes in layers above the critical thickness. There are three main criteria that a strain-relief buffer structure must satisfy to be successful. First, the amount of strain relief in the structure must be controlled, i.e., it must be predictable. Second, the density of dislocations emanating from the top of the buffer layer and threading through the device structure must be low, and third, the surface must be sufficiently planar to allow a well-defined device structure to be manufactured on top. In this paper, we describe the models and experimental observations of these three parameters.

Relaxation of Constant Composition Layers

We shall only consider in detail the case where an

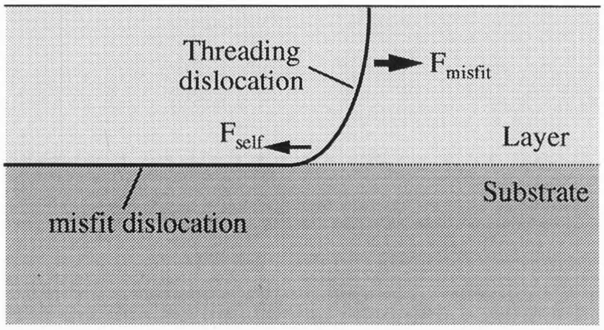

Figure 1. Lateral movement of a threading dislocation to produce a misfit dislocation in the interface between layer and substrate.

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layer has a fixed composition. Equilibrium models of strain relief of such single constant composition layers generally assume that a planar array of dislocations is present at the interface between a planar layer and the substrate. The strain energy of the layer is then calculated and the minimum energy found for a given misfit dislocation density, giving the equilibrium strain of the layer as a function of film thickness (e.g., Matthews and Blakeslee, 1974; Willis *et al.*, 1990; Fitzgerald, 1991; Jain *et al.*, 1993a,b). Another way of performing the calculation is to find the strain at which a glissile threading dislocation becomes unable to overcome the line-tension of the misfit dislocation it leaves behind at the interface (e.g., Matthews *et al.*, 1970; Fitzgerald, 1991) (Figure 1). The force on the dislocation due to the strain in the layer is found by integrating the force per unit length on the dislocation over its length, from interface to surface. Although there are several variations of this calculation, corresponding to more or less complicated and realistic situations, they all produce an equation of the form:

$$\varepsilon = \{(A_0/h) \ln(B_0 h)\} \qquad (1)$$

where h is the layer thickness and A_0 and B_0 are constants, determined by the elastic constants of the material and the dislocations' Burgers vectors, and are roughly equal to 0.2 nm and 0.2 nm^{-1} respectively for 60° dislocations in (001) layers. Because the log term varies only slowly, the strain in the layer decreases roughly as 1/h. Also, equation (1) is independent of the original misfit, f; once relaxation begins, the amount of strain in the layer is expected to depend only on the film thickness h. The critical thickness, h_c, occurs when the equilibrium strain in the layer is equal to the misfit, f, i.e.,

$$h_c = \{(A_0/f) \ln(B_0 h_c)\}. \qquad (2)$$

Experimentally, it is found that high misfits tend to produce island growth rather than planar growth, resulting in a very high threading dislocation density (> 10^8 cm^{-2}), which only slowly decreases as the film thickness increases. Lower misfit layers, however, have a regular array of 60° dislocations (an example is shown in Figure 2) but with significant deviations from the ideal array used to derive equation (1). For example, the spacing of the array is not regular due to the lack of dislocation mobility in the interface and dislocations are often seen lying parallel to, but above the interface. We have grown many constant composition layers, mainly in the $In_xGa_{1-x}As/GaAs$ system, over a range of x from 0.05 to 0.25 and a range of h from 40 nm to 3 μm, using chemical beam epitaxy (CBE), atomic-layer molecular beam epitaxy (ALMBE) and MBE. The strain in these layers, measured by double-crystal X-ray diffraction (DCXRD) is shown in Figure 3, together with equation (1), marked equilibrium. Poor quality layers, i.e., those with a very high threading dislocation density (> 10^8 cm^{-2}) or very large surface roughness etc. have been excluded. Generally, they exhibit even less relaxation than those included in Figure 3. There is a very obvious discrepancy between theory and experiment, with all significantly relaxed layers having about an order of magnitude more strain than predicted by the simple equilibrium models, i.e., they are "metastable."

It has been proposed that metastability is due mainly to "kinetic effects," i.e., finite dislocation velocities (e.g., Dodson and Tsao, 1987, 1988; Tuppen and Gibbings, 1990), since a layer does not have enough time to relax during growth before the material is cooled, "freezing-in" the dislocations and preventing further relaxation. To investigate how stable these structures are, we have performed several annealing experiments on samples with both small, intermediate and large amounts of relaxation. Surprisingly, very little change in strain was observed for any of the samples (usually less than 5% relaxation), unless the annealing temperature was high enough to allow significant amounts of dislocation climb (Lourenço *et al.*, 1994). So, even though the layers are not in equilibrium, they do appear to be stable for most practical purposes. This implies that the finite dislocation velocity plays little part in metastability in this system.

It has also been suggested that the relaxation process is a result of heterogeneous dislocation multiplication sources, i.e., particulates (e.g., Fitzgerald *et al.*, 1989; Gibbings *et al.*, 1989), small stacking faults (Eaglesham *et al.*, 1989), etc.. However, it does not seem likely that these effects can account for the observed behaviour. Particulate contamination of the original substrate surface would be expected to change the misfit dislocation density in layers above the critical thickness, but

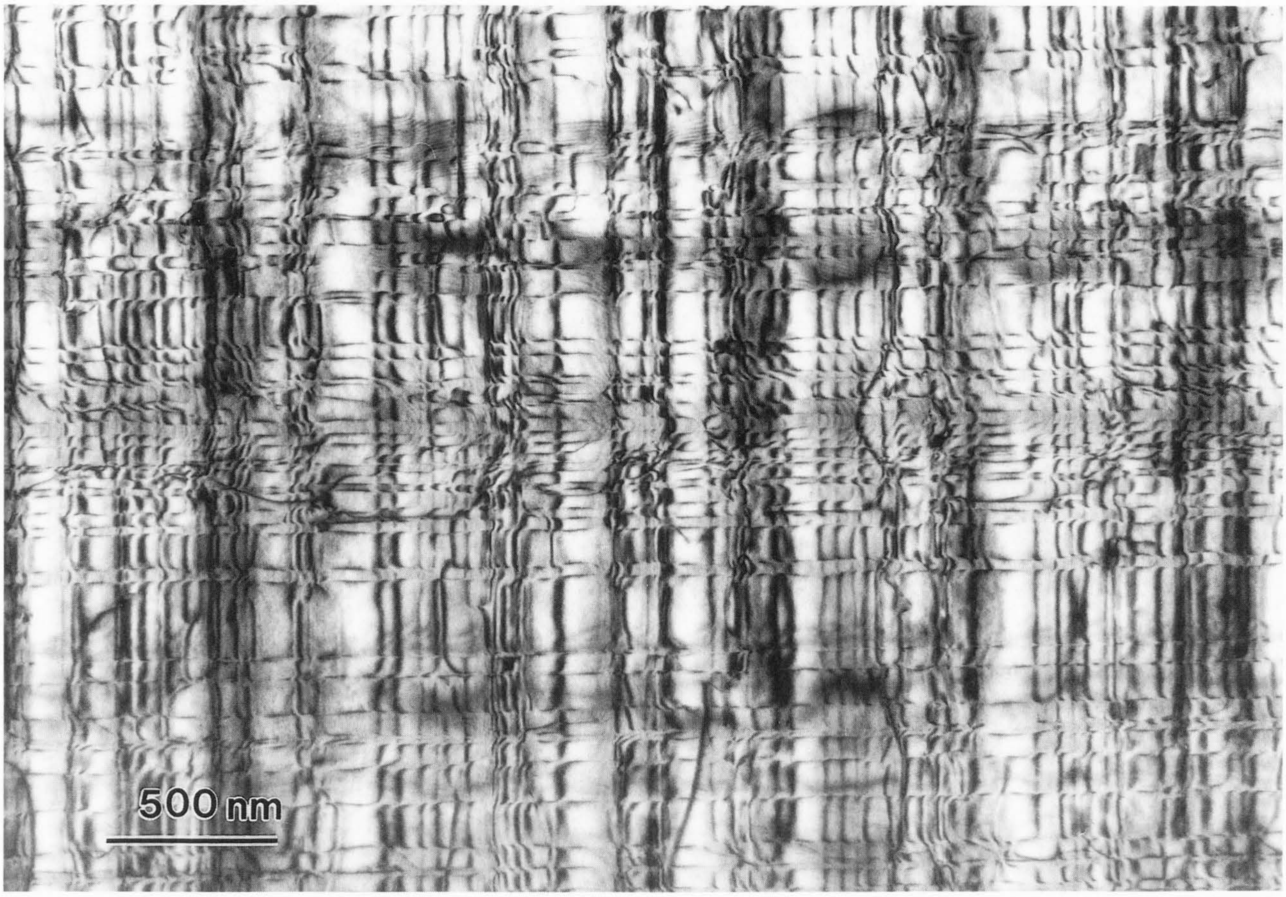

Figure 2. Plan-view transmission electron micrograph of a 60° misfit dislocation array between a constant composition 200 nm thick $In_{0.14}Ga_{0.86}As$ layer and a (001) GaAs substrate.

would not be expected to give the two phases of misfit dislocation introduction that are observed, the second being at thicknesses about an order of magnitude larger than h_c. This is borne out by the limited area experiments of Fitzgerald *et al.* (1989) and the X-ray topography and DCXRD study of Dugdale *et al.* (1993). In this latter study, it was found that substrates of different dislocation densities relaxed to different extents for thicknesses slightly above the critical thickness, but thick layers showed the same relaxation, irrespective of the substrate quality. The data agree very well with that presented in Figure 3.

Heterogeneous nucleation sources incorporated during growth are also not consistent with the observed behaviour. Since the force on a dislocation in a strained layer is proportional to its length, the critical stress required to produce a misfit dislocation from a heterogeneous source is inversely proportional to its size. Since the strain in layers which are above the critical thickness but which have not relaxed significantly is essentially the same as in layers below the critical thickness, the only way that heterogeneous sources can produce the observed behaviour is if they gradually increase in size as the layer thickens. Furthermore, their increase in size must depend upon the misfit strain, since the thickness at which relaxation starts is inversely proportional to the strain. This seems unlikely behaviour for particulate contamination or stacking fault loops. It should also be mentioned that plan-view transmission electron microscopy (TEM) was performed on all the samples represented in Figure 3 and no evidence of sources of this type was found. The predicted equilibrium strain is also shown in Figure 3 together with the criteria for dislocation blocking and multiplication.

The key to understanding the observed behaviour lies in understanding the underlying processes of dislocation generation and movement. Two processes in particular seem to have great importance; a) dislocation "blocking," or work hardening, and b) dislocation multiplication. Dislocation blocking (Freund, 1990a,b) occurs when a threading dislocation (dislocation A) moving in the layer (and leaving a misfit dislocation behind in the interface as it does so), encounters a pre-existing dislocation in its path (dislocation B; Figure 4). Since

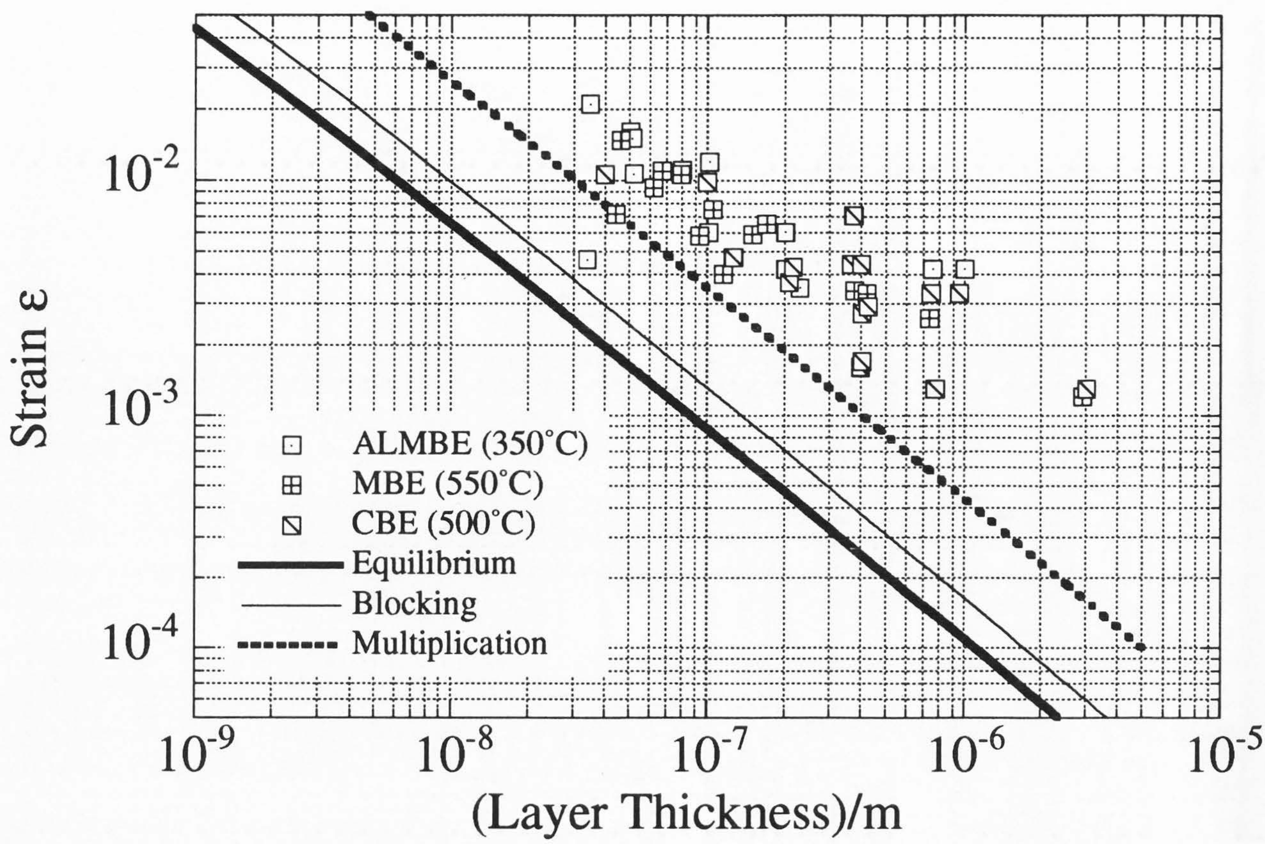

Figure 3. Experimental DCXRD measurements of strain in constant composition $In_xGa_{1-x}As$ layers on (001) GaAs.

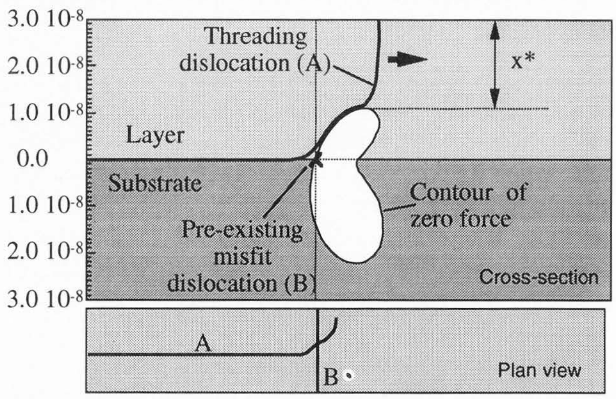

Figure 4. Schematic illustration of dislocation blocking. The mobile dislocation (A) is forced to bow round the pre-existing dislocation (B) due to the interaction of their stress fields.

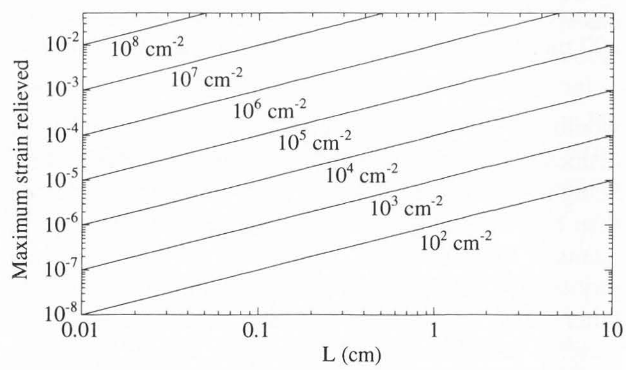

Figure 5. The maximum strain that can be relieved in a square substrate, with sides of length L, from a limited density of dislocation sources (10^2 cm^{-2} to 10^8 cm^{-2}).

a portion of a glissile dislocation comes to rest at the point where the force it experiences is zero, dislocation A will be forced to bow around the strain field of dislocation B if dislocation B is fixed. The path that dislocation A takes is simply the line where the force on dislocation A due to dislocation B exactly cancels the force on dislocation A due to the misfit stress (Figure 4). Dislocation A can only pass dislocation B if the net force on the remaining mobile segment, of length x*, is sufficient to overcome the line tension. This is exactly similar to the force-balance model of critical thickness and residual strain, except that the stress field of dislocation B should be included. Freund (1990a) made the approximation that x* must be greater than the critical

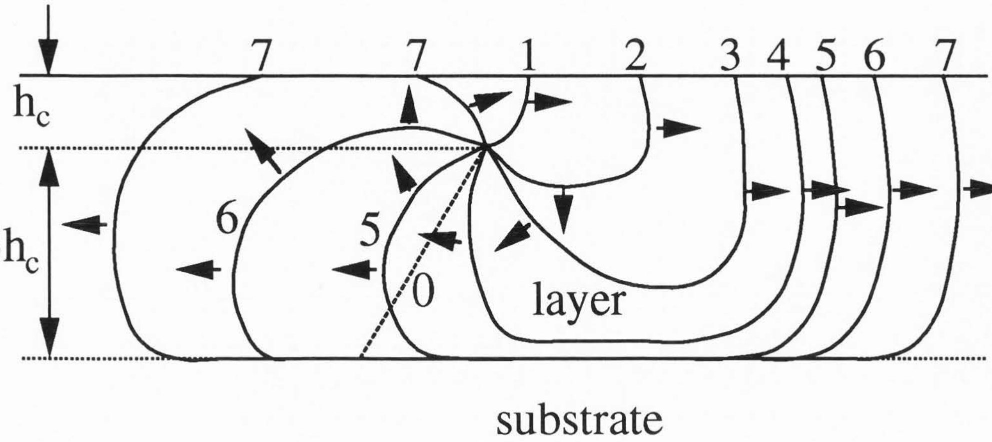

Figure 6. Schematic illustration of a spiral multiplication source in a strained layer. A dislocation, (0), cross-slips (1), and then expands (2, 3, 4, 5, 6), until a new half-loop is formed (7) and the original configuration (1) is recovered.

thickness, h_c; the resulting criterion for blocking can be approximated to an equation similar in form to (1), i.e.,

$$\varepsilon = \{(A_b/h) \ln(B_b h)\} \qquad (3)$$

where A_b is roughly equal to $1.5A_0$ to $2A_0$ and $B_b \approx B_0$.

More rigorous calculations are possible, but often analytical solutions cannot be obtained. However, all calculations where an analytic solution can be obtained can be approximated to an equation of the form of (3), but with slightly different values for the constants A_b and B_b. It is easy to see that blocking will always occur before equilibrium strain is reached; dislocation blocking thus provides one mechanism which prevents strained layers from reaching equilibrium. However, as can be seen from Figure 3, the effect is not strong enough to explain the relatively large strain in single layers. The second mechanism which we consider here is dislocation multiplication. It is well-known that even without dislocation blocking, the initial number of mobile threading dislocations is too small to account for the observed amount of relaxation (Matthews *et al.*, 1970; Fitzgerald, 1989; Fitzgerald *et al.*, 1989; Beanland, 1992). This effect is dependent upon the substrate size, since the length of misfit dislocation that can be produced by a moving threading dislocation depends upon the distance that it can travel. Figure 5 shows the maximum strain that can be relieved for a given substrate size, assuming that all threading dislocations propagate to the edges of a square substrate. Since the typical threading dislocation density of a GaAs wafer is below 10^5 cm^{-2}, and the wafer size of typically 1-5 cm, it is clear that new dislocations must be generated to relieve even 0.1% strain. The most feasible dislocation multiplication mechanisms in strained layers appear to be of the Frank-Read and spiral type (Beanland, 1992); an example of a spiral source is shown in Figure 6 at various stages of operation. The pinning point, about which the mobile threading dislocation rotates, has not been identified to date; however, the most likely possibility is that the pinning point is simply another dislocation (Beanland, 1995).

A transmission electron micrograph of dislocation loops in a constant composition layer of $In_{0.15}Ga_{0.85}As$, with a configuration consistent with the operation of such a source, is shown in Figure 7. It is straight-forward to show that, if the dislocation cannot pass through the layer-substrate interface, the minimum thickness for such sources to operate is about $4h_c$ (Beanland, 1992). However, these mechanisms are not **likely** to operate until the layer thickness is substantially greater than this, since the pinning point about which the dislocation circulates must lie further than about $3h_c$ from the interface and h_c below the surface [we should note here that several observations of these types of sources have shown dislocation pile-ups extending deep into the substrate (Lefebvre *et al.*, 1991; LeGoues *et al.*, 1991); this appears to be possible only because of the large stress generated by a pile-up and should not change the thickness at which multiplication starts].

The criterion for multiplication to occur can once more be expressed in a form similar to (1) i.e.,

$$\varepsilon \geq \{(A_m/h) \ln(B_m h)\} \qquad (4)$$

where $A_m \approx 4A_0$ and $B_m = B_0$. This correlates very well with the observed onset of relaxation, as shown in Figure 3. Thus, the onset of relaxation appears to be related to the onset of dislocation multiplication. Misfit dislocation multiplication also has a very important side effect; the generation of new threading dislocations. In (001) layers, the threading dislocation density, ρ_{TD}, is related to the misfit dislocation density, ρ_{MD}, by the equation (Matthews *et al.*, 1970; Fitzgerald, 1989):

$$\rho_{TD} \geq \{(2n\rho_{MD})/L\} \qquad (5)$$

where L is the mean misfit dislocation length, and n is the mean number of threading dislocations associated with a misfit dislocation, which can vary from zero

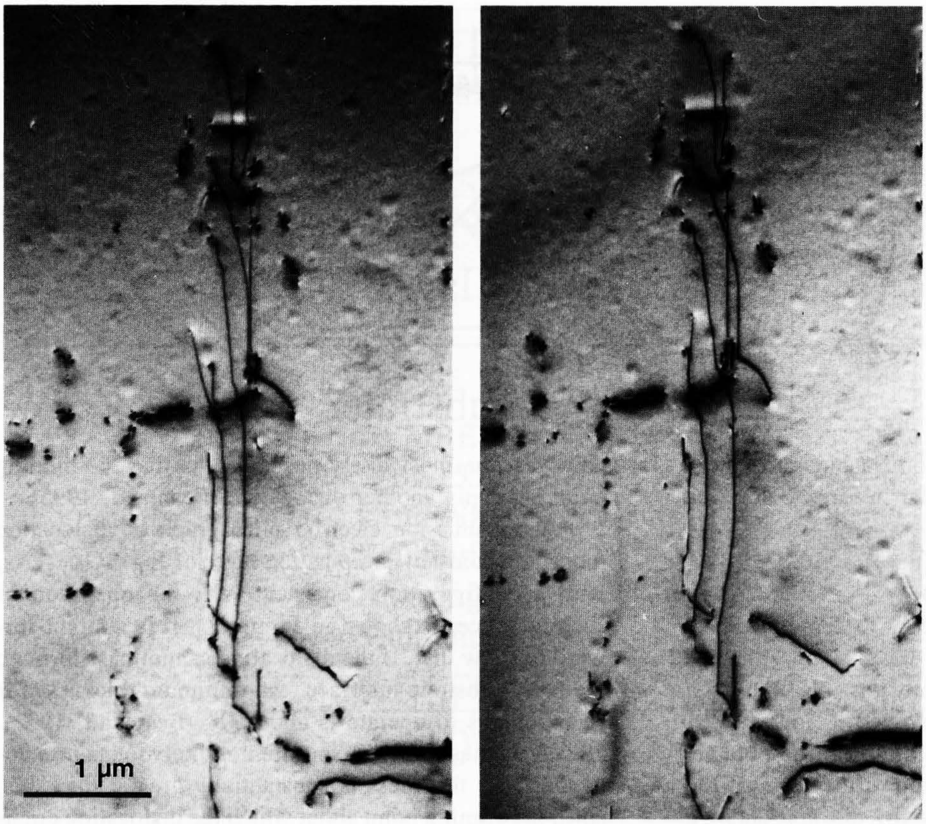

Figure 7. Plan-view stereo-pair transmission electron micrographs of a dislocation source in a 800 nm thick $In_{0.15}Ga_{0.85}As$ strained layer; several dislocation loops can be seen to lie on the same {111} plane emanating from a single point.

(both ends terminate at the wafer edges or the back of the substrate) to two (both ends terminate at the layer surface). When multiplication occurs, each new misfit dislocation must have two new threading dislocations associated with it (n = 2). We observe an increase of threading dislocation density from below 105 cm^{-2} at the initial stages of misfit relief to the high 10^7 cm^{-2} in layers several μm thick for all samples except those with very low misfits ($x \leq 0.05$). This very rapid rise in ρ_{TD} indicates that the mean misfit dislocation length L decreases rapidly with increasing thickness. This is correlated with the appearance of pile-ups of dislocations in the layer resulting from the repeated operation of multiplication sources; an example of such a pile-up in a $In_{0.17}Ga_{0.83}As$/GaAs layer 370 nm thick is shown in Figure 8. These pile-ups must provide very efficient blocking of any mobile dislocations moving perpendicular to them, resulting in shorter and shorter misfit dislocation lengths as relaxation proceeds and a rapid rise in the threading dislocation density.

It is well-known that a distinctive cross-hatch pattern, aligned with <110> directions, appears on the surface of strained layers during relaxation. This pattern consists of triangular ridges with rounded tops separated by V-shaped grooves; a plot of the surface height as a function of position, taken from an atomic force microscope (AFM) image of such a surface, is shown in Figure 9. These ridges appear on the surface of the layer some time after the appearance of the first dislocations, and appear to be closely linked with dislocations lying parallel to and above the interface such as those shown in Figure 8. In layers with strain-thickness characteristics below the multiplication criterion, only fine random roughness is observed. Ridges begin to form when the strain-thickness characteristics touch the curve for dislocation multiplication, and rapidly increase in height and number, until the whole surface is covered by ridges, such as those shown in Figure 9. TEM investigation shows that dislocations lying parallel to but above the interface have a similar spacing to the ridges, and bending of the TEM foil, consistent with a ridge lying above these dislocations, can also be seen. Furthermore, the distribution of dislocations above the interface is asymmetric, with those parallel to [110] being far more common than those parallel to $[\bar{1}10]$; this asymmetry is also observed in the ridge pattern on the surface (Beanland *et al.*, 1995). The most likely explanation of these ridges appears to be a local enhancement of growth rate above the dislocation due to their strain fields. Once more, this indicates the importance of dislocation multiplication mechanisms in the relaxation of strained layers (Beanland and Boyd, 1995).

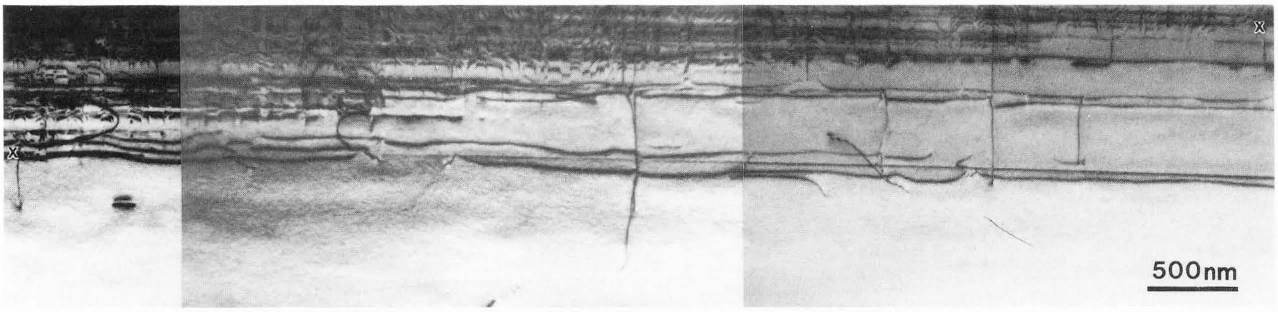

Figure 8. Plan-view TEM photomontage of a dislocation source and pile-up in a 370 nm thick $In_{0.17}Ga_{0.83}As$ strained layer. The interface between layer and substrate intersects the foil surface along x-x; the interfacial dislocation array can be seen above this line. The dislocation source and pile-up lie completely inside the epilayer, and the dislocations produced by this source could be followed for over 200 μm.

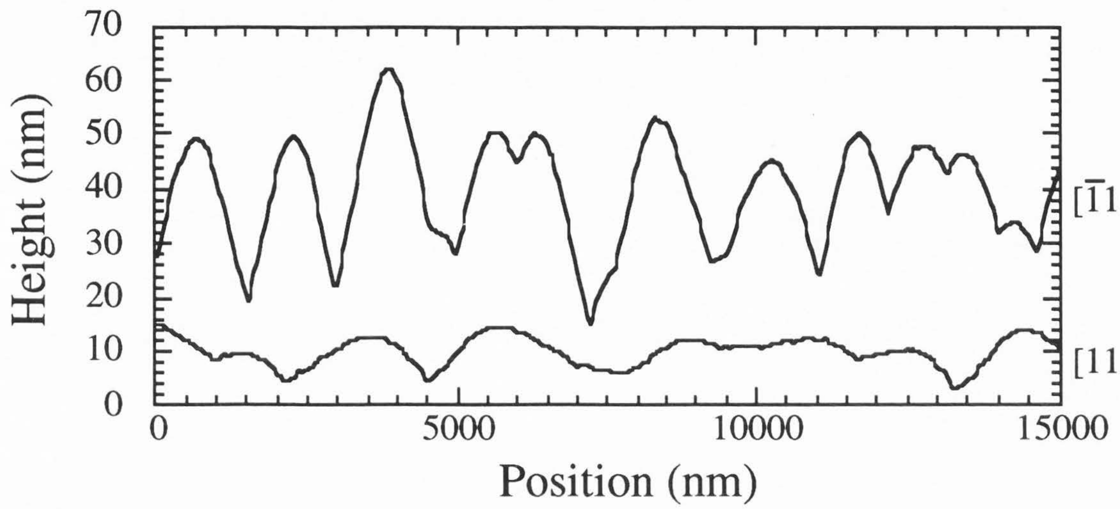

Figure 9. Surface height as a function of position taken from a AFM image of the surface of a 370 nm thick $In_{0.17}Ga_{0.83}As$ strained layer (the same as shown in Figure 8). The surface consists of triangular ridges with rounded tops about 1 μm wide and up to 40 nm high, separated by V-shaped grooves. Ridges lying parallel to [$\bar{1}$10] are much higher than those lying parallel to [110].

Summary and Future Prospects

Relaxation in low-misfit, constant composition $In_xGa_{1-x}As$ layers on GaAs is reproducible (and hence predictable) although it does not follow equilibrium models. Although dislocations appear close to the equilibrium critical thickness, significant relaxation does not occur until layers exceed several times the critical thickness. Relaxation appears to be limited by the availability of new dislocations (i.e., dislocation multiplication), and is independent of the growth technique and the growth temperature. Dislocation blocking plays an important part by immobilising existing threading dislocations, leading to high dislocation densities in layers more than ~ 1 μm in thickness. Surface striations appear at a similar thickness to the onset of significant relaxation.

Because of the high threading dislocation density that develops after any significant amount of strain is relieved ($> \sim 0.5\%$), constant composition layers are not suitable candidates for strain-relief buffer layers, despite the reproducibility of relaxation. However, other structures may be more suitable.

It is hoped to extend this approach to layers in which the misfit strain varies with thickness. These layers should exhibit quite different behaviour. In particular, linearly graded layers are well-known to have virtually complete relaxation and very low threading dislocation densities. The strain in such layers is expected to be independent of thickness; it will be interesting to see if the blocking and multiplication criteria have a similar lack of dependence on layer thickness.

It is unclear how to produce a flat surface if inhomogeneous strain fields are responsible. One way of influencing the strength of the pattern is to change the density of dislocations which lie parallel to the surface, possibly by using "barrier" layers of different material. Low-temperature growth has shown some success in reducing the strength of the cross-hatch pattern (e.g., Howard *et al.*, 1992). However, it is possible that the pattern may reappear during growth of an overlying device layer.

Acknowledgements

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Discussion with Reviewers

D.C. Houghton: The time-temperature combinations used for anneals should be defined more completely. The kinetics of misfit dislocation injection and propagation (time-temperature-effective stress) have been treated in the literature for the Ge_xSi_{1-x}/Si system. A comparison with the present experimental data from $In_xGa_{1-x}As$/GaAs would be illuminating, since no dependence on thermal history in the latter is apparent in this work.

Reviewer III: The discussion of the connection (or lack thereof) between metastability and "kinetic effects" is interesting, but too qualitative in light of the experiments of Bean and co-workers, Nix and co-workers and others on SiGe/Si systems which show significant relaxation during a post-growth anneal. Perhaps a little more quantitative information on annealing temperatures, time at temperature, etc. could be included?

Authors: Only preliminary results of the annealing experiments have been published to date (Lourenço *et al.*, 1994), and it is perhaps premature to include them in this paper. Further experiments are in progress investigating the time-temperature-effective stress behaviour of $In_xGa_{1-x}As/GaAs$ structures. However, it seems clear even from the data presented in this paper and the initial annealing experiments that relaxation in the $In_xGa_{1-x}As$/GaAs system is not determined by growth temperature, and that annealing at temperatures below 850°C does not induce relaxation. The main point is that relaxation of $In_xGa_{1-x}As$ layers on GaAs is not determined by "kinetic" effects, as can be the case in Ge_xSi_{1-x} on Si. It is also worth noting that the dislocation velocity in $In_xGa_{1-x}As$ is typically two, or more, orders of magnitude larger than in Ge_xSi_{1-x}, and so kinetic effects might be expected to play a much smaller role in strained $In_xGa_{1-x}As$ layers.

D.C. Houghton: Figure 3 should have the extent of strain relaxation displayed in addition to the nominal (fully strained) thickness data. This plot is meaningless as it stands since no information on the degree of strain relaxation is available.

Authors: All models of strain relief predict that the strain in a layer above the critical thickness should be independent of the original misfit. Since all the samples described in Figure 3 are above the critical thickness, and hence partially relaxed, they should lie on the "Equilibrium" curve, irrespective of the misfit strain or the amount of relief that has occurred. We agree that it is useful to examine series in which the misfit strain is kept constant, allowing the amount of strain relieved to be easily seen; however, the main point which Figure 3 seeks to make is that there is a minimum strain that a layer relaxes to which is far above equilibrium, and that this minimum strain follows a law similar in form to the equilibrium equation.

D.C. Houghton: Figure 5 should cover the range of strains observable in TEM and XRD and important for strained layer devices, ie 10^{-4} to 10^{-2}.

Authors: The range of strains mentioned is covered in Figure 5. To restrict the strain **only** to this range would mean that typical substrate threading dislocation densities and substrate sizes would not be shown. The incredibly low strains which are shown indicate the distinct lack of strain relief that will occur with a very low dislocation source density.

D.C. Houghton: The final sentence in the **Relaxation of Constant Composition Layers** section should be modified since there is little evidence to support a correlation between multiplication and the occurrence of surface ridges.

Authors: We had hoped that the description of the evolution of the cross-hatch pattern would show the correlation between the onset of significant relaxation and the onset of surface roughening. We believe that the relaxation behaviour of $In_xGa_{1-x}As/GaAs$ layers is most consistent with dislocation multiplication, and not dislocation nucleation. We have recently performed *in-situ* laser-light-scattering studies of the surface roughness of CBE-grown $In_xGa_{1-x}As$ layers, and have found that the cross-hatch pattern starts to form some time after the critical thickness has been passed, and at a thickness which coincides exactly with the onset of significant relaxation of the layers, which implies that the onset of the cross-hatch pattern is in some way associated with dislocation multiplication (Beanland and Boyd, 1995). Experiments are also under way to investigate the origin of the cross-hatch pattern.

Reviewer III: In the first paragraph in the **Relaxation of Constant Composition Layers** section, tradition is followed in the introduction of the "equilibrium" approach and the "energy" approach to a critical thickness criterion. It is noted that these different approaches lead to a result in the form of equation (1). In fact, it has been demonstrated that for a system of certain characteristics, the two approaches are different forms of the **same** result, and they lead to precisely the **same** thickness-mismatch equation [cf. Freund LB (1992) Dislocation mechanisms of relaxation in strained epitaxial films. MRS Bulletin **17**, 52-54]

Authors: We agree with the referee's comments on the equivalence of the "force-balance" and the "energy balance" approaches to obtaining the critical thickness. However, there is a whole family of critical thickness calculations, depending upon the degree of realism required, for example, the effect of a range of dislocation spacings rather than a constant one, the effect of the surface step left in the wake of a travelling threading dislocation, or the emphasis placed on the core energy or cancellation of surface tractions. There is, of course, only one true critical thickness for a given strained layer and all models are only approximations to the real situation. The main point is that all the variations give essentially the same answer (equation 1), as they should, with slight differences in the constants A_0 and B_0.

Reviewer III: On the question of the blocking mechanism of Freund (1990b), it might be noted that one conclusion from that work is that it is unlikely that blocking can play a significant role in the relaxation process unless the films are thin in some sense. It is interesting to see the prediction based on the blocking mechanism included here, but perhaps it should not be surprising that it does not emerge as a dominant mechanism.

Authors: It does not seem clear to us that the blocking mechanism proposed by Freund (1990b) can only occur in "thin" layers, although this is, indeed, stated explicitly by Freund (1990b). All the calculations which Freund (1990b) performed depend only on the strain-thickness product, not the absolute thickness. This seems to indicate that exactly the same mechanisms can operate in thick layers with low strain as in thin layers with high strain, or, to phrase it another way, the mechanism depends only on the effective stress as defined by Dodson and Tsao (1987, 1988). Blocking would thus be expected to occur in partially relaxed layers as well as essentially unrelaxed layers, although it would act as a limit to relaxation, rather than preventing it completely.

Reviewer III: I would urge the authors to label the most important features of significance as clearly as possible and to describe in somewhat more detail what is being shown. The pinned point shown schematically in Figure 6 is essential to the whole argument. How does such a barrier come to be; what is it physically?

Authors: It is rather difficult to label features on a stereo pair, since the letters "float" around when the two pictures are viewed in stereo. When Figure 7 is viewed in stereo, it is clear that several half-loops are present which lie on the same {111} glide plane, with some segments which have cross-slipped on to the alternate {111} glide plane. Unfortunately, the origin of these loops was lost in the thinning process, and only half of the dislocation loops produced can be seen in the Figure. However, from the configuration of the loops it is clear that they have emanated from some point source. It cannot be stated categorically that the source is a pinning point, rather than a contaminant particle; however, no evidence of such particles was seen, although many dislocations lying above the interface have been seen in this, and indeed every, significantly relaxed layer. We now believe that the pinning points which allow spiral and Frank-Read sources to operate are simply other dislocations. It turns out that one in four reactions between dislocations in a layer can produce pinning points and a suitable configuration for multiplication to take place.

Reviewer III: I wonder if the surface waviness shown in Figure 9 could be due to stress driven surface diffusion, or is there other evidence that it is due to the influence of the non-uniform strain of the atom attachment process?

Authors: We have no fixed opinion on the origin of the cross-hatch pattern. Stress-driven surface diffusion of already deposited material or a difference in the incorporation rate due to strain both seem likely, or unlikely, at present, although it must be admitted that a difference in growth rate solely due to the strain in the material does not seem particularly likely, since the growth rate of strained and unstrained material is the same, as far as we know. It is also possible that geometrical factors, such as large surface steps produced by pile-ups of dislocations, may have a large part to play.